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THESIS

THE INFLUENCE OF THERMOMECHANICAL PROCESSING PARAMETERS ON THE ELEVATED TEMPERATURE MECHANICAL BEHAVIOR OF A 6061 ALUMINUM - ALUMINA METAL MATRIX COMPOSITE MATERIALS

By

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December 1990

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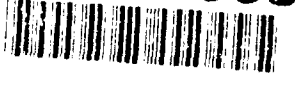

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The Influence of Thermomechanical Processing on the Elevated Temperature Mechanical Behavior of 6061 Aluminum - Alumina Metal Matrix Composite Materials

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Submitted in partial fulfillment of the requirements for the degree of

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I. INTRODUCTION

There is considerable interest in discontinuous Metal Matrix Composites (MMC's) both in industry and in the Department of Defense (DOD). This is due to the unique physical and mechanical characteristics and potential life-cycle cost savings when these materials are included in engineering structures [Ref. 1]. These composite materials offer significant property advantages which may be utilized in strong, light-weight components to provide improved performance of naval ships, aircraft, missile systems and armor protection systems [Ref. 2].

MMC's can be designed to meet the requirements of specific tasks [Ref. 3]. Advantageous properties and characteristics include:

- high stiffness-to-density ratios;
- increased fatigue and wear resistance;
- high strength-to-density ratios;
- controlled thermal expansion characteristics;
- elevated temperature resistance;
- damping properties; and
- corrosion resistance.

Fiber reinforced composites offer the highest specific stiffness along the direction of reinforcement allignment, while particle-reinforced composites are more nearly isotropic in their properties. The discontinuous MMC's are also easier to process via casting or powder metallurgy routes.

One of the main current drawbacks is the lack of an extensive data base applicable to the response of MMC's in a variety of conditions. Some other problem areas are:

- heat treatments response;
- residual stresses; and
- adverse reinforcement-matrix interactions.

All of these topics still require extensive research and analysis to determine the nature of the processes involved.

An MMC is a macroscopic combination of a metallic matrix and ceramic reinforcement. The reinforcement may be a continuous in the form of a fiber or discontinuous in the form of whiskers or particles [Ref. 4]. Typical reinforcements used include graphite, silicon carbide (SiC), alumina (Al_2O_3), boron carbide (B_4C) and boron. These ceramic materials have very high melting points (up to 2000°C). They are generally insoluble in the matrix and thus provide a very hard phase to be imbedded in the matrix. This results in advantages in, for example, abrasion resistance for use in applications such as piston cylinders [Ref. 5], aircraft components and tooling [Ref. 6].

The matrix aluminum alloys most often used in MMC's are 2124, 7475 and 6061 due to their age hardening properties and strength potential. The aluminum alloy used here is 6061, and the ceramic reinforcement is alumina in particulate form. In many respects, this composite is similar to conventional aluminum alloys in that it is nearly isotropic and can be processed and formed by many traditional metalworking techniques. In discontinuous MMC's the

properties of the matrix play a significant role in the strength of the composite [Ref. 7].

Powder metallurgy methods of composite fabrication have received a great deal of attention in the past, but current fabrication techniques utilizing casting methods have provided a new, equally viable option for consolidation.

MMC's using discontinuous fiber reinforcements are of interest to designers because they often have advantages in cost and manufacturing procedures are simpler when compared to continuous fiber MMC's. Particulate reinforcements result in the most isotropic material when compared to the continuous or discontinuous fiber MMC's. This allows more flexibility in design and a similar approach to those method employed with current structural metals.

The production of good quality, cast composites by Dural Aluminum Composites Corporation has made it possible to investigate a wider range of the mechanical properties of cast Discontinuous Metal Matrix Composites (DMMC's). The purpose of this thesis is to investigate the effect of thermomechanical processing on the mechanical properties of an Al 6061-Alumina MMC, particularly the strength and ductility of the composite. Also, the dependence of such properties on test temperatures is of prime concern here.

II. BACKGROUND

A. HISTORY

Support for research into metal matrix composites has fluctuated considerably over the past 30 years (Figure 1). The number of funded programs peaked between 1968 and 1970 and then declined in the 1970's, only to increase again during the 1980's. In the 1960's the DOD began research on metal matrix composite materials with a view to their use in armor systems to defeat small arms fire and rocket-assisted projectiles, i.e., shape charges. This was during the Vietnam war era. Then, research on metal matrix composites focused mainly on consolidation and evaluation of materials and less on post-consolidation processing and control of microstructural development. Some composite properties such as density may be predicted by means of the "rule of mixtures". Other structure-sensitive properties, such as strength and ductility, were less well understood. At that time, discontinuous MMC's were fabricated utilizing powder metallurgy methods. Fine ceramic powders were blended with coarse metallic powders and typically the resultant compact would be sintered and extruded. Often, the ceramic and metal powder particles were of substantially different size, resulting in non-uniform reinforcement distribution and erratic mechanical properties. Thus, during the 1970's there was a decrease in composite materials research, both because production

of good quality materials was not technologically feasible at that time and because the war was ending.

In the 1980's, there was a resurgence of interest in metal matrix composite materials for aerospace and automotive applications. This reflected improved fabrication techniques and better understanding of the processing. Potential military applications have expanded to include advanced systems which are classified and not included in Figure 1 [Ref. 8]. This research has expanded as new techniques in fabrication and processing have been developed. Of particular interest to the current research is the advent of ingot (casting) methods to produce discontinuous metal matrix composites (DMMC's). Currently, composites are designed primarily to satisfy specific engineering requirements for service performance. One way to further improve the utility of these materials is to enhance their ductility or even to make them superplastic at elevated temperatures. Superplastic response has been observed in many materials including several aluminum alloys but little data is available on this phenomenon in composite materials.

Research on thermomechanical processing (TMP) techniques at the Naval Postgraduate School has considered various TMP schemes to enhance the subsequent ductility of the processed material. The previous research work on Al-Mg alloys has obtained superplastic elongations in excess of 1000% at 300°C [Ref. 9]. These rolling schedules were modified and employed to process the Al 6061-Alumina composite. The ambient temperature mechanical properties were then

examined [Ref. 10]. Essentially, this research extends the previous investigation to consider the elevated temperature strength and ductility as influenced by the processing.

B. THE MATRIX ALLOY

DMMC properties depend on numerous factors, such as the type of reinforcement, the size, the distribution and the matrix characteristics. The matrix should effectively transmit the load to the reinforcement and should resist or stop crack propagation [Ref. 5]. Aluminum and its alloys are versatile metallic materials for engineering applications. The alloy designated as 6061 constitutes the matrix of the composite of interest here. It is heat treatable, of low density and good specific strength [Ref. 5]. The matrix generally is the component which limits the service temperature of the composite. The composition limits for 6061 are listed in Table I below [Ref. 11].

Table I. COMPOSITION LIMITS FOR 6061 ALUMINUM (IN WEIGHT PERCENT)

Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Misc
0.8	0.7	0.4	0.15	0.12	0.35	0.25	0.15	0.05

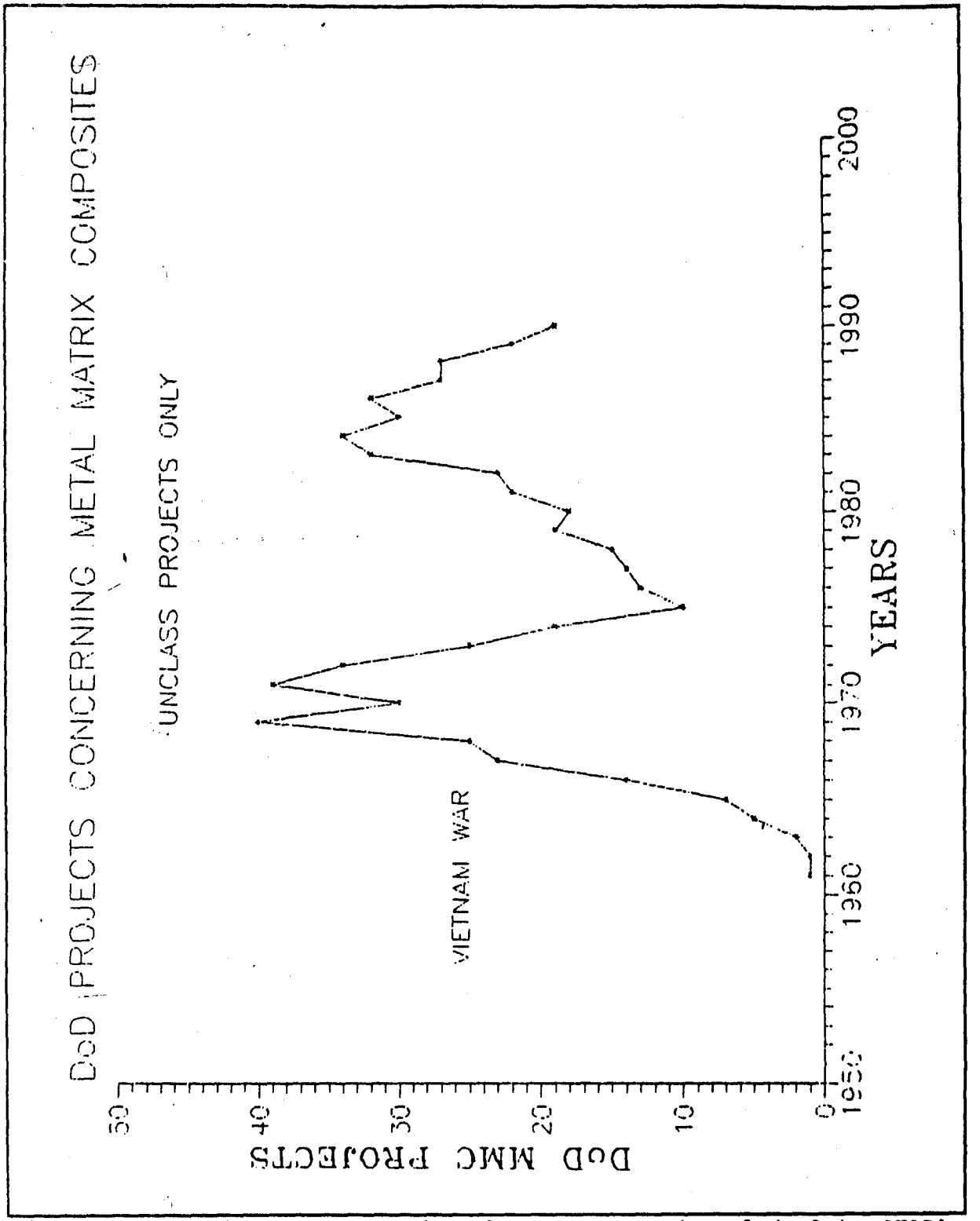

Figure 1 Funded programs by the government related to MMC's from 1960 to present. Only unclassified projects were considered.

C. THE REINFORCEMENT

In Aluminum-based, cast DMMC materials, particles of the reinforcement are added to the molten aluminum prior to casting. This requires some form of pretreatment to assure subsequent bonding between the matrix and ceramic particles. The matrix must wet the particles to facilitate such bonding. Indeed, alumina fibers have been reported to exhibit good wetting characteristics with aluminum [Ref. 12]. Good bonding allows effective distribution of loads to the reinforcement and therefore improves stiffness and the strength of the composite. The ceramic material used here was alumina in particulate form, which was cast into the aluminum 6061 matrix using a proprietary technique developed by Dural Composites Corporation of San Diego California. The methods utilized avoid problems with the bonding at the matrix to ceramic interface [Ref. 13]; this results in a composite of good metallurgical integrity. The alumina was in the form of a fine particulate, of irregular shape and with sharp corners. The particle size corresponds to an approximate diameter of 10μm. Volume fractions of 10 and 15 percent were obtained.

The strength of the ceramic is very high and is retained at temperatures up to 1500°C. Thus, reinforcing alumina represents a hard, non-deforming particle in the Al6061 matrix for all temperatures up to the melting point of the matrix. The melting range of the 6061 alloy is 580-650°C while the melting point of the alumina is about 2000°C. The coefficient of thermal expansion for

the aluminum is nearly five times greater than that for alumina: 14.1 X $10^{-6}K^{-1}$ for aluminum versus 3.4 X $10^{-6}K^{-1}$ for alumina.

D. COMPOSITES

The DMMC properties are affected by three main factors: the reinforcement particles, the matrix, and the interface between the particles and the matrix. Cast composites typically exhibit random orientation in at least two dimensions due to the preforming processes [Ref. 5]. Small additions of ceramic particles or fibers provide major improvements in adhesive wear of the composite.

The grain size during casting is determined by the nucleation rate as well as the flow of fluid during solidification. The nucleation rate is influenced by the rate of cooling and by nucleation sites. However, most commercial reinforcements do not act as a heterogeneous nucleation catalysts for Aluminum alloys [Ref. 14]. Instead, during solidification, the reinforcement acts as a barrier to solute diffusion ahead of liquid/solid interface and hence the last portion of the metal to solidify will be close to the reinforcement/matrix interface.

E. PROCESSING OF METAL MATRIX COMPOSITES

To date, there has been only limited research on thermomechanical processing of cast DMMC's. A uniform reinforcement distribution is essential to the effective load carrying capacity of the composite. Non-uniform distribution of the ceramic will be take the form of stringers or clusters often in

association with porosity. Together, these factors lower the ductility, strength and toughness of the composite. It is generally thought that the non-uniform distribution of particles in the cast MMC is produced during solidification and is not the result of failure to achieve good mixing during melting. Particles are pushed by growing dendrite arms into the last regions to freeze during solidification [Ref. 15]. Subsequent processing, however, can damage the reinforcement particles as well as improve their distribution. Most DMMC are given a homogenization treatment prior working to minimize the solute concentration gradients in the composite. Cast DMMC also display smaller secondary dendrite arm spacing than unreinforced alloys. After annealing for homogenization most DMMC are mechanically worked. Extrusion is often used to improve the strength and ductility of the composite by improving the dispersion of the reinforcement. Also, hot rolling of SiC in 6061 has been shown to improve both strength and ductility [Ref. 16].

III. EXPERIMENTAL PROCEDURE

A. MATERIALS AND SECTIONING

The material used was a DMMC with a matrix consisting of 6061 Aluminum alloy and with either 10 or 15 volume percent alumina (Al_2O_3) particles added. The nominal particle size was approximately 10μm. The material was received in an as-extruded condition as well as in the as-cast condition from Dural Aluminum Composites Corporation of San Diego Ca.

All of the MMC originated from cast material. Most of the provided material had been subjected to a 17:1 extrusion [Ref. 20], corresponding to a true strain of $\epsilon = 2.83$. The extruded material was received in the form of rectangular bars of cross section measuring 76mm X 19mm (2.9in X 0.75in), with rounded edges, and approximately 0.5m (19.7in) in length. The sectioning into billets approximately 45mm X 35mm X 19mm (1.77in X 1.4in X 0.75in) for rolling was done manually with a hacksaw. The sides were trimmed square to help avoid edge cracking during rolling. The as-cast material was provided in the form of discs 125mm (5 inches) in diameter and 13mm (0.5 inch) thick which had been sectioned from large casting.

B. THERMOMECHANICAL PROCESSING

Solution treatment for 90 minutes at 560°C was accomplished for microstructural homogenization of the matrix utilizing a

Lindberg type B-6 Heavy Duty furnace. The billets were immediately quenched in a water bath after solution treatment.

Billets were then placed in a Blue M furnace, model 8655F-3, to provide heating at the rolling temperature for 30 minutes prior to the first rolling pass. Each billet was thus able to equilibrate at the desired rolling temperature.

Billets were rolled utilizing a Fenn Laboratory Rolling mill with a rolling schedule summarized in Table II. This schedule is similar to that developed here at NPS for processing of superplastic Al-Mg [Ref. 9] alloys and the same as used for previous work on this composite [Ref. 10]. During the rolling

Table II ROLLING SCHEDULE

ROLLING PASS NO.	ROLL CHANGE (.08IN-.01IN)	MILL GAP (IN)	% STRAIN (PER PASS)
1	+(8+0)	0.64	14.6
2	-(1+2)	0.54	15.6
3	-(1+2)	0.44	18.5
4	-(1+2)	0.34	22.7
5	-(1+2)	0.24	29.4
6	-(0+6)	0.18	25.0
7	-(0+6)	0.12	33.3
8	-(0+5)	0.07	41.6
9	-(0+2.2)	0.048	31.4

process, a silicone spray lubricant was used from the fifth rolling pass onward to eliminate sticking of the composite during rolling. After each rolling pass the billets were returned to the furnace for 30 minutes of reheating and annealing. At the completion of the final rolling pass the billets were quenched in water to ambient temperature.

C. MACHINING

Coupons of the rolled MMC's were machined for tensile testing to dimensions summarized in Figure 2. The machining problems experienced in earlier work [Ref. 10] using carbide-tipped tools in the endmills were resolved by using diamond-tipped tools. This allowed machining of coupons with no warping or distortion and provided smooth surfaces and a very low rejection rate of samples.

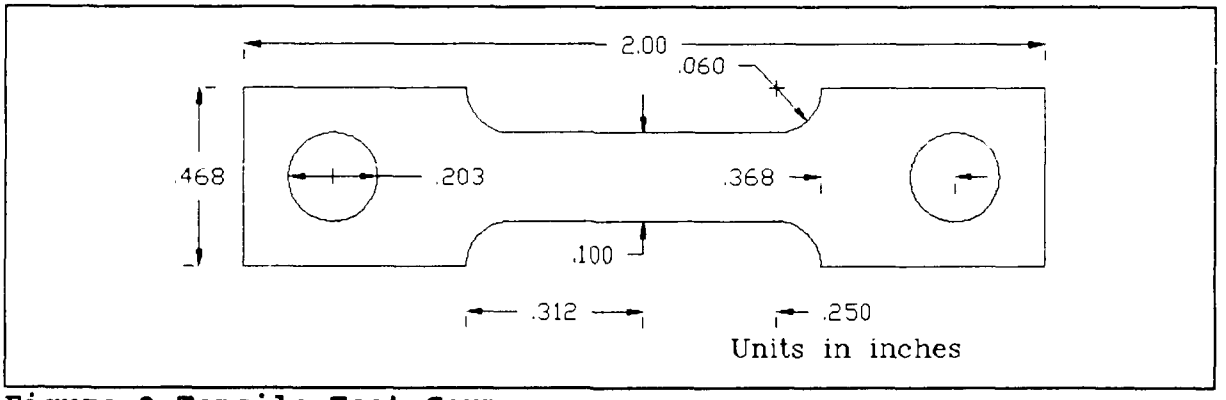

Figure 2 Tensile Test Coupon

D. TENSILE TESTING

Tensile testing was performed on an Instron Model 6027 testing machine. An Astro tubular furnace (1200°C, 60Hz, 110V) was mounted on the Instron and provided temperature control with an accuracy of

±2°C. Output signals from the Instron test machine were provided simultaneously to a Hewlett Packard Model 3852A Data Aquisition Unit and a Hewlett Packard strip chart recorder. This provided both load vs. time records from both systems and stress vs. strain plots for each sample following computer manipulation.

Five thermocouples were mounted in the furnace to provide temperature control and to maintain temperature distribution. The thermocouples were distributed over a length from five inches above and to five inches below the sample.

Testing was conducted utilizing crosshead speeds varying from 0.05mm/min to 254 mm/min, providing strain rates from $6.7 \times 10^{-5} S^{-1}$ to $6.3 \times 10^{-1} S^{-1}$. The test temperatures ranged from 200°C to 500°C and a single test at 550°C was also accomplished. Samples were equilibrated and held at temperature for 5 min. prior to each test and each test was at a temperature within 10°C of desired test temperature.

E. DATA REDUCTION

Engineering stress vs. strain curves were obtained via computer programs using standard methods. Subsequent plots of yield strength, tensile strength and ductility vs. strain rate and vs. temperature were produced for both 10 and 15 volume percent MMC materials.

F. OPTICAL MICROSCOPY

A Zeiss ICM-405 Optical Microscope was utilized for the optical microscopy. Samples were mounted and polished using procedures similar to those of previous work [Ref. 10]. It was found that polishing times outlined in Table III are the maximum times most likely to be used if polishing is accomplished by hand. The goal was to achieve a white-appearing background matrix with the grey alumina particles. Too little polishing resulted in too many scratches. If samples are overpolished, the smaller alumina particles appear to have pulled out and embedded in the matrix, giving the effect of a much more higher alumina volume percentage.

Table III SAMPLE PREPARATION PROCEDURES

STEP #	POLISHING MEDIUM	TIME	COMMENTS
1	320 Grit	2 min.	light pressure
2	400 Grit	2 min.	light pressure
3	600 Grit	2 min.	light pressure
4	6 micron diamond paste	3 min.	light pressure
5	3 micron diamond paste	2 min.	light pressure
6	1 micron collodial silica	1-2 min.	light pressure

IV. RESULTS

A. REINFORCEMENT DISTRIBUTION

The as-cast microstructure is inhomogeneous (Figure 3(a)). This is understood to be the result of the sweeping of the reinforcement particles into interdendritic regions during solidification following casting.

Hot extrusion of a casting to the form of a round cornered rectangular bar, utilizing a reduction of 17:1, homogenized the particle distribution. It can be seen from figure 3(b) that the particle distribution is banded and clustering of particles is still evident.

Further improvement in particle distribution was accomplished during hot rolling utilizing a reheating temperature of 500°C. However, an even greater degree of homogenization was obtained upon warm rolling at 350°C in place of the hot rolling at the higher temperature. This is especially evident in the improved extent to which the clustering resulting from solidification has been eliminated (Figure 3(c) and 3(d)).

B. THE TEMPERATURE DEPENDENCE OF THE STRENGTH

The addition of particles results in strengthening at ambient temperatures [Ref. 7, 10, 11]. Further, the strength of both of the warm rolled materials is higher than that of the extruded material when testing is conducted at ambient temperature [Ref. 10]. At temperatures above 200°C, the warm rolled

composites become weaker than the material experiencing extrusion only (Figure 4). Also, the warm rolled composites appear to be even weaker than the unreinforced Al-6061 matrix material. This may be attributed to the formation in the warm rolled composites of a refined structure with boundaries of sufficient misorientation to support accelerated recovery processes, including possibly boundary sliding mechanisms, during deformation (Figure 4).

Comparison of the strength versus temperature data of this research (Figures 5-8) reveals that increased volume fraction of particles results in increased composite strength at test temperatures below 350°C (for the same rolling conditions). However, above 400°C, strengths tend to be the same for both composite volume fractions, irrespective of rolling temperature.

C. TEMPERATURE DEPENDENCE OF THE DUCTILITY

Ductility data obtained in this research is presented in Figures 9-12. Ductility versus temperature data for the 10 vol. pct. alumina material, rolled at 350°C or 500°C, are contained in Figures 9 and 10, respectively. Similar data for the 15 vol. pct. material are in Figures 11 and 12. In general, the 10 vol. pct. composite exhibits somewhat higher ductility than the 15 vol. pct. material for the same processing history and testing conditions. This may be seen by comparison of Figures 9 and 11 or Figures 10 and 12. For all combinations of composite composition and processing history, the ductility decreases with increasing strain rate for test temperatures up to about 350°C. Above this test

temperature, an inversion in behavior is seen and ductility appears to increase with increasing strain rate. This effect is most prooounced for the 10 vol. pct. composite rolled at a temperature of 350°C. This material appears to exhibit a maximum in ductility of approximately 130 pct. at a test temperature of 400°C and strain rate of 6.7×10^{-2} s^{-1}.

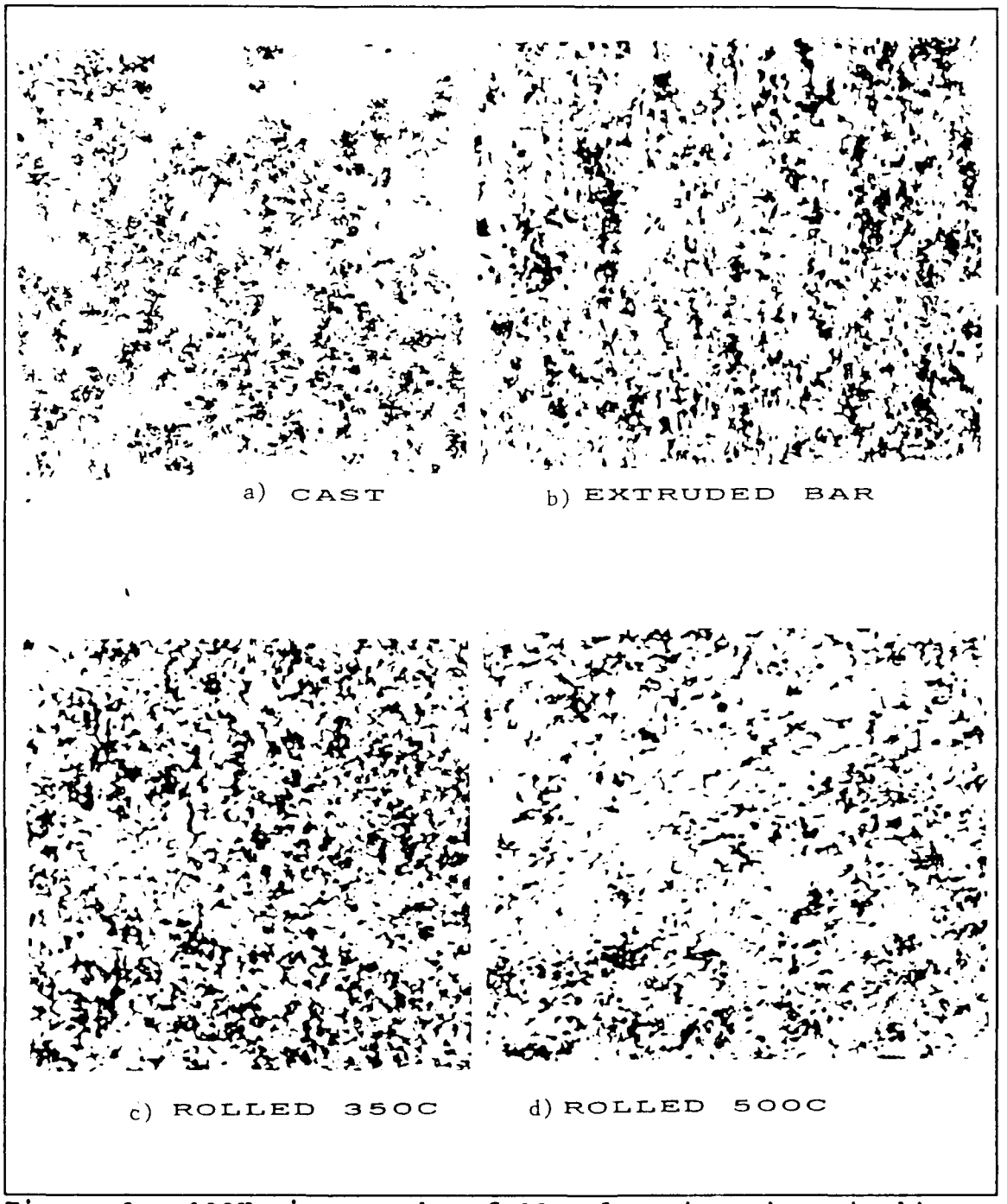

Figure 3 100X micrographs of 10 vol. pct. a) cast b) extruded c) rolled at 350°C d) rolled at 500°C

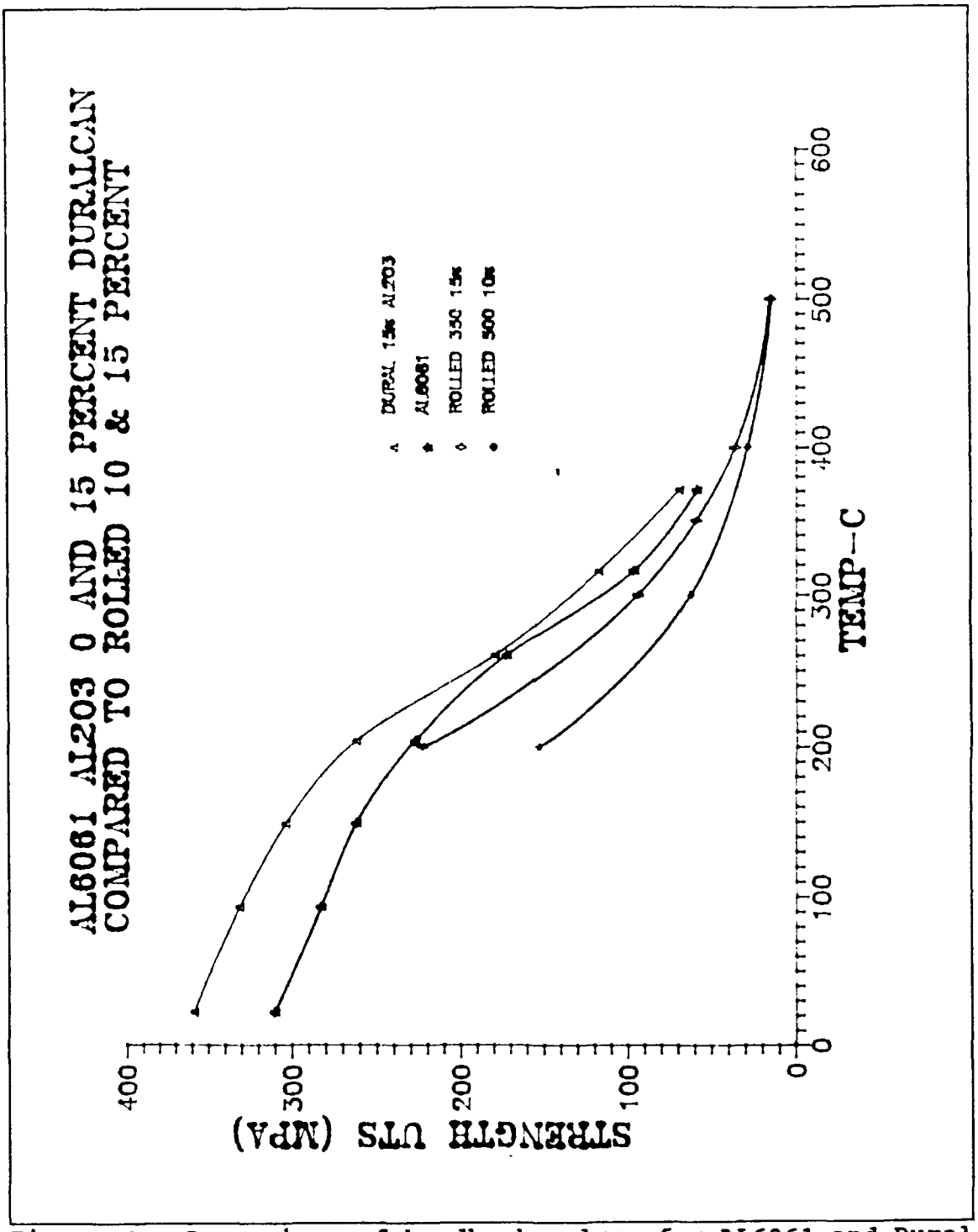

Figure 4 Comparison of handbook values for AL6061 and Dural data on 15 vol. pct. MMC compared to material which was rolled at tested in this experiment.

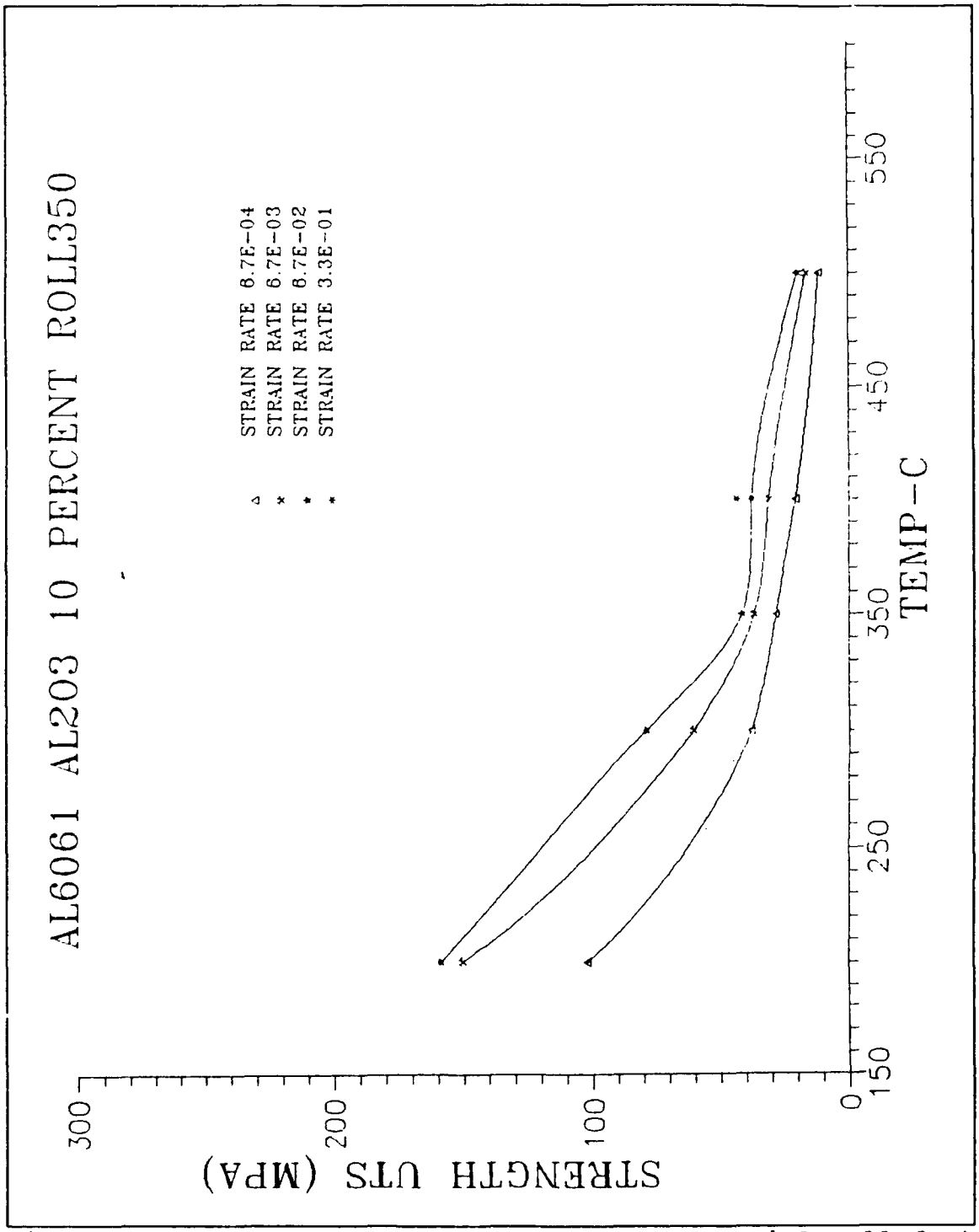

Figure 5 **Strength data for 10 vol. pct. material rolled at 350°C**

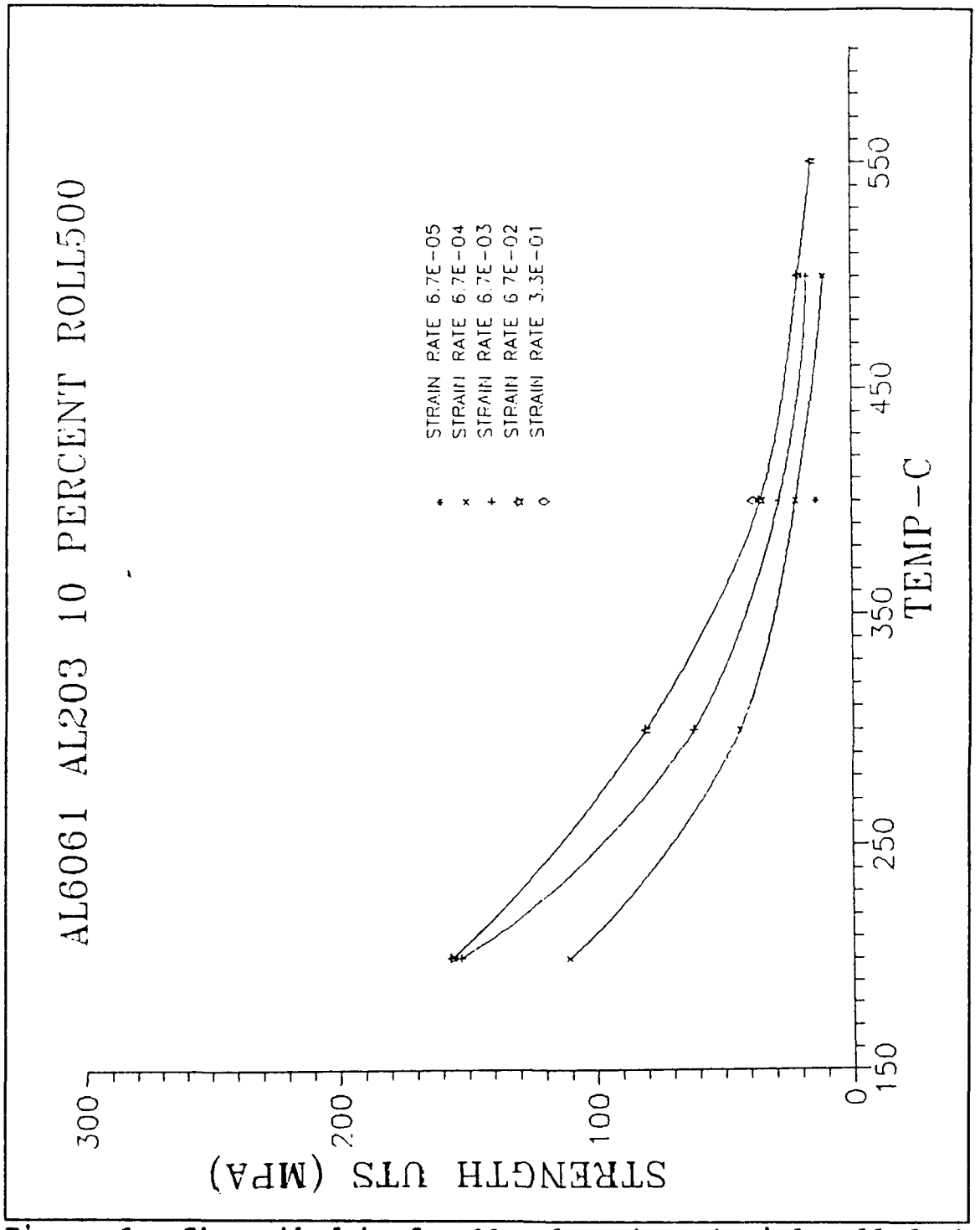

Figure 6 **Strength data for 10 vol. pct. material rolled at 500°C**

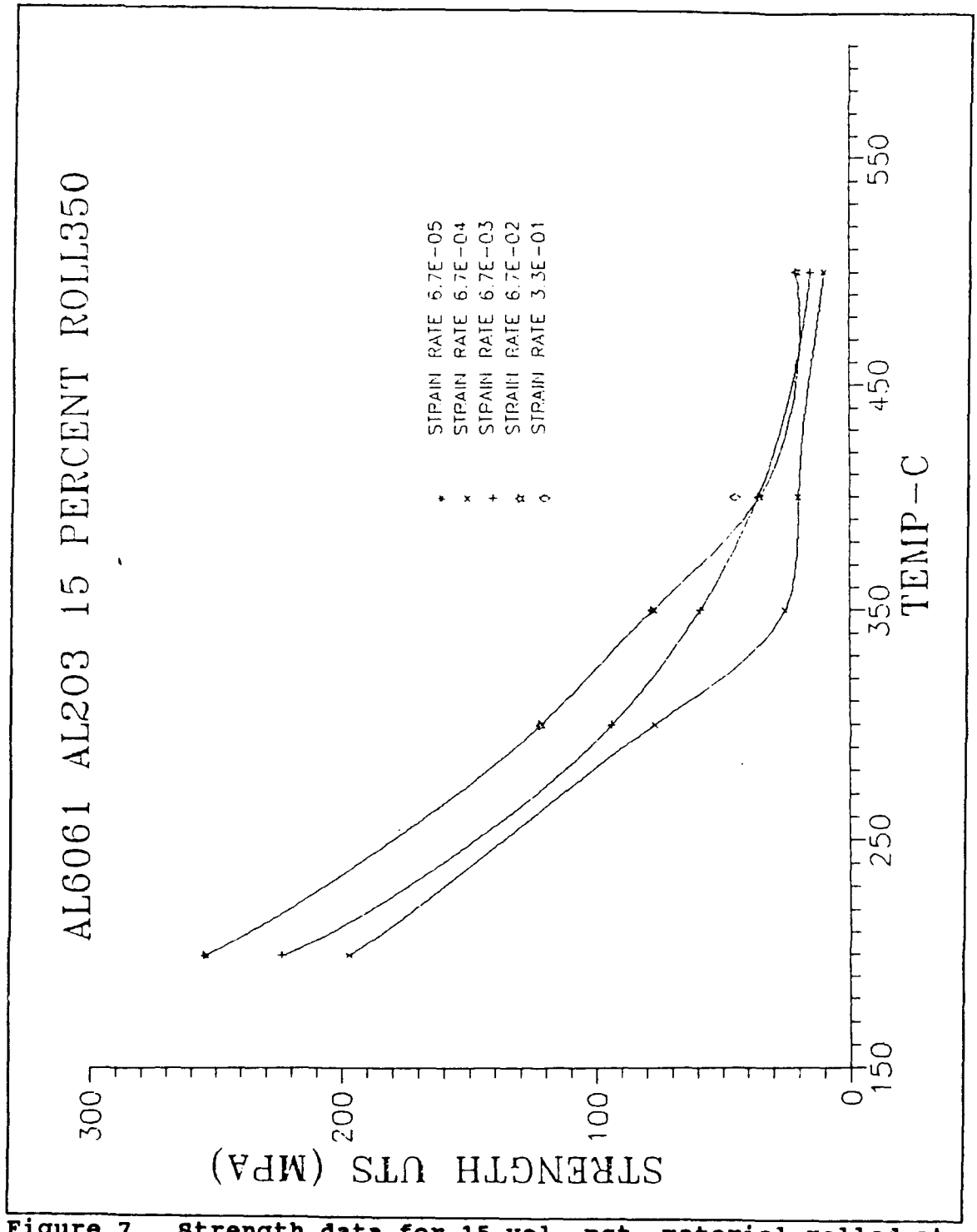

Figure 7 **Strength data for 15 vol. pct. material rolled at 350°C**

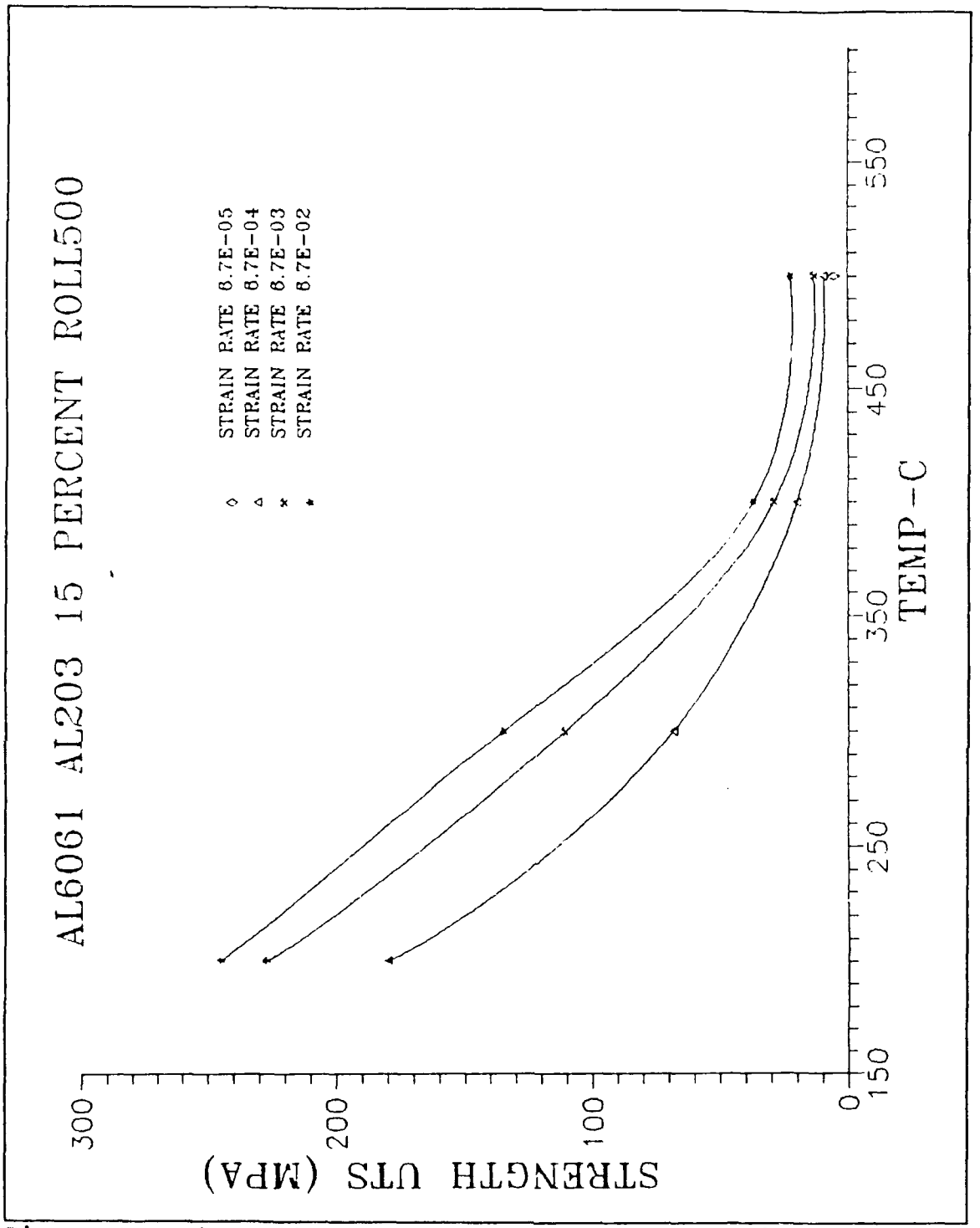

Figure 8 **Strength data for 15 vol. pct. material rolled at 500°C**

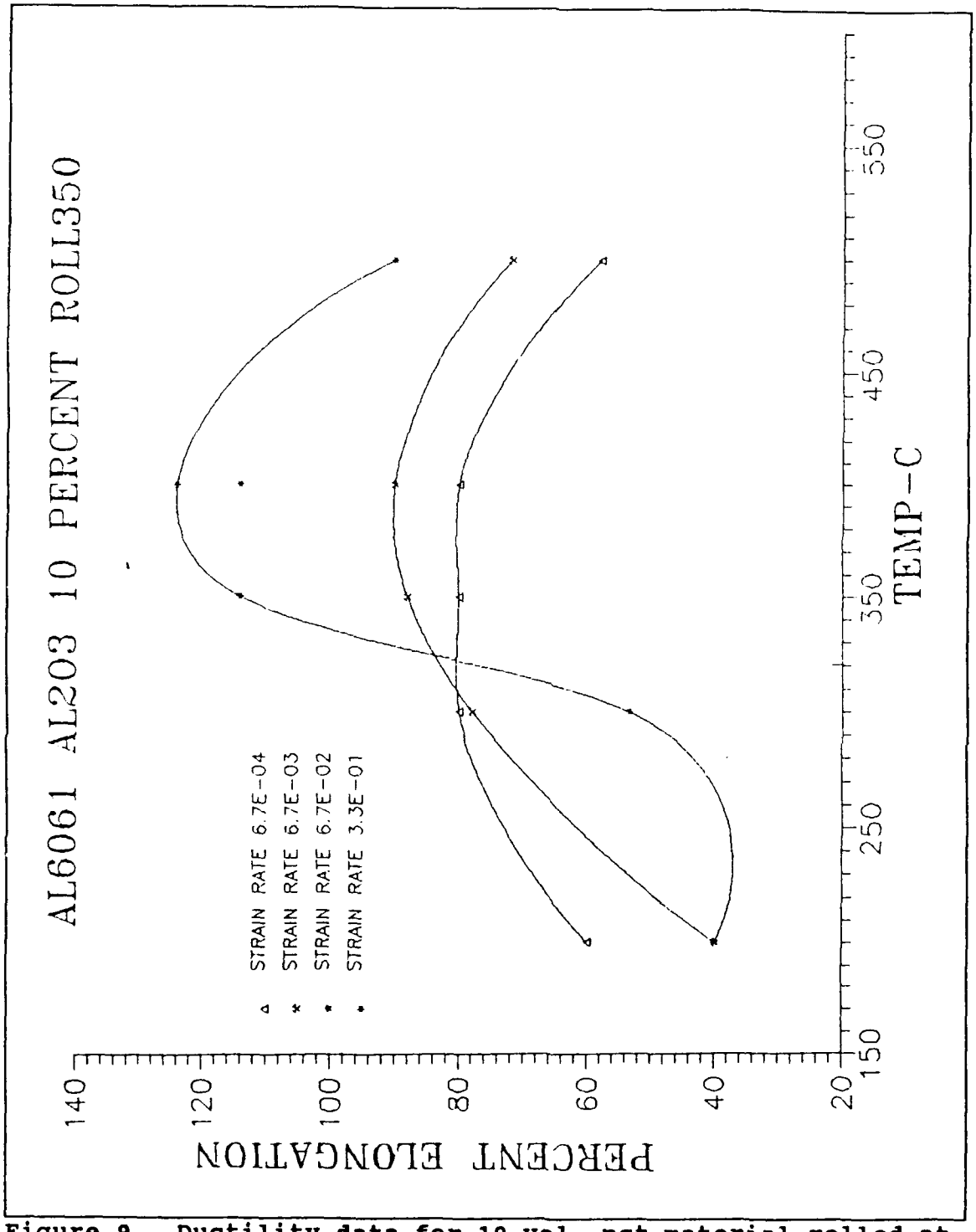

Figure 9 Ductility data for 10 vol. pct material rolled at 350°C

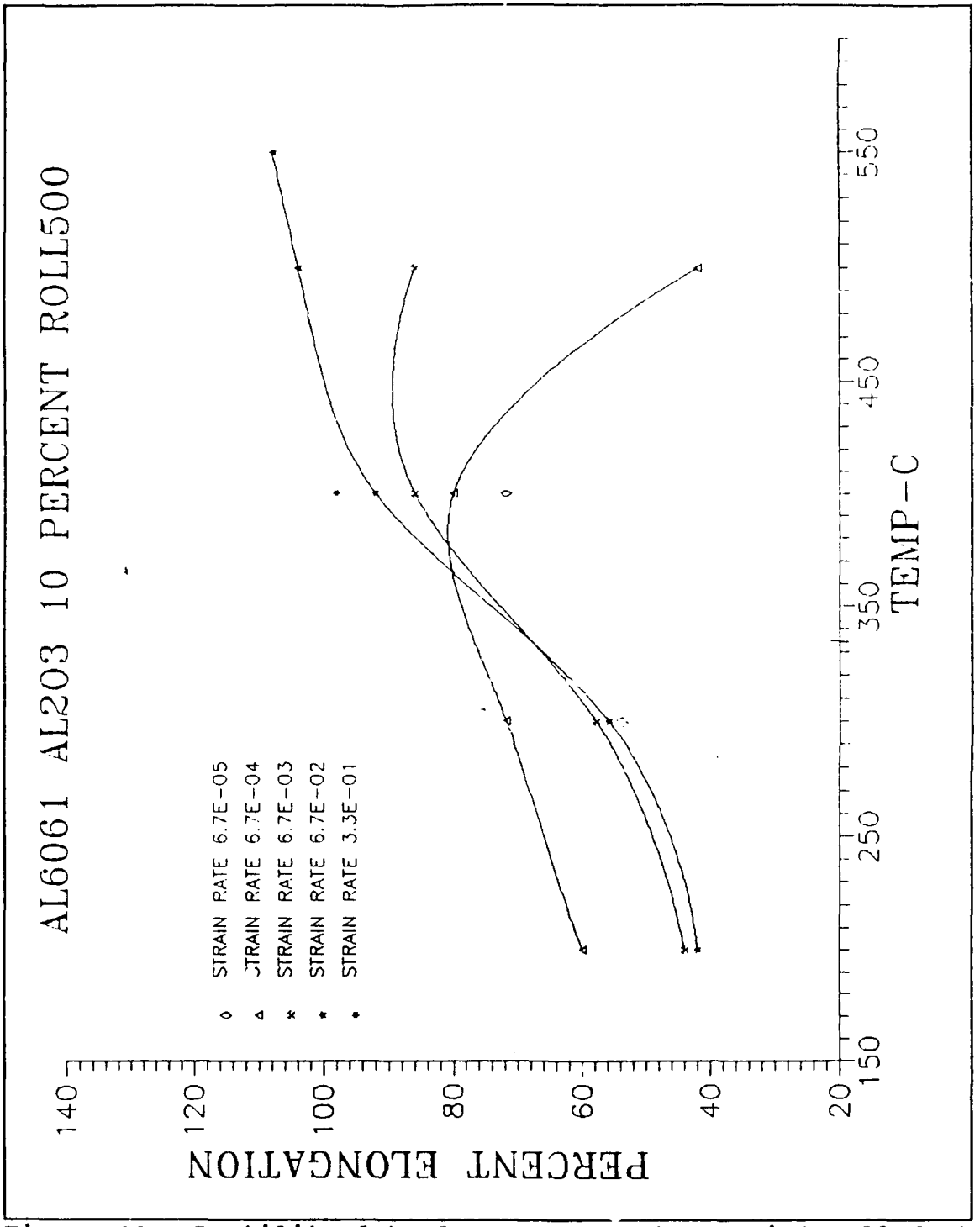

Figure 10 **Ductility data for 10 vol. pct material rolled at 500°C**

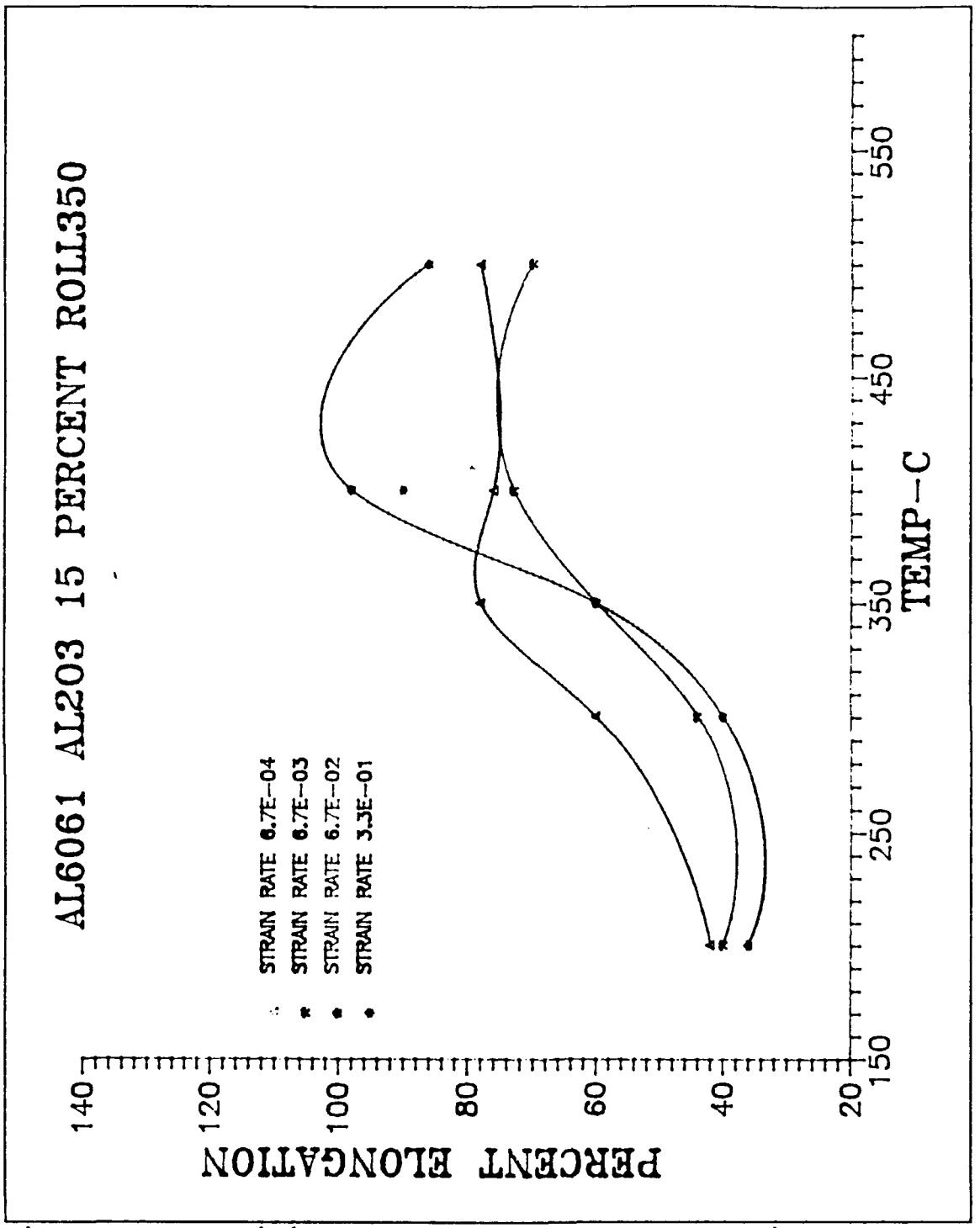

Figure 11 Ductility data for 15 vol. pct. material rolled at 350°C

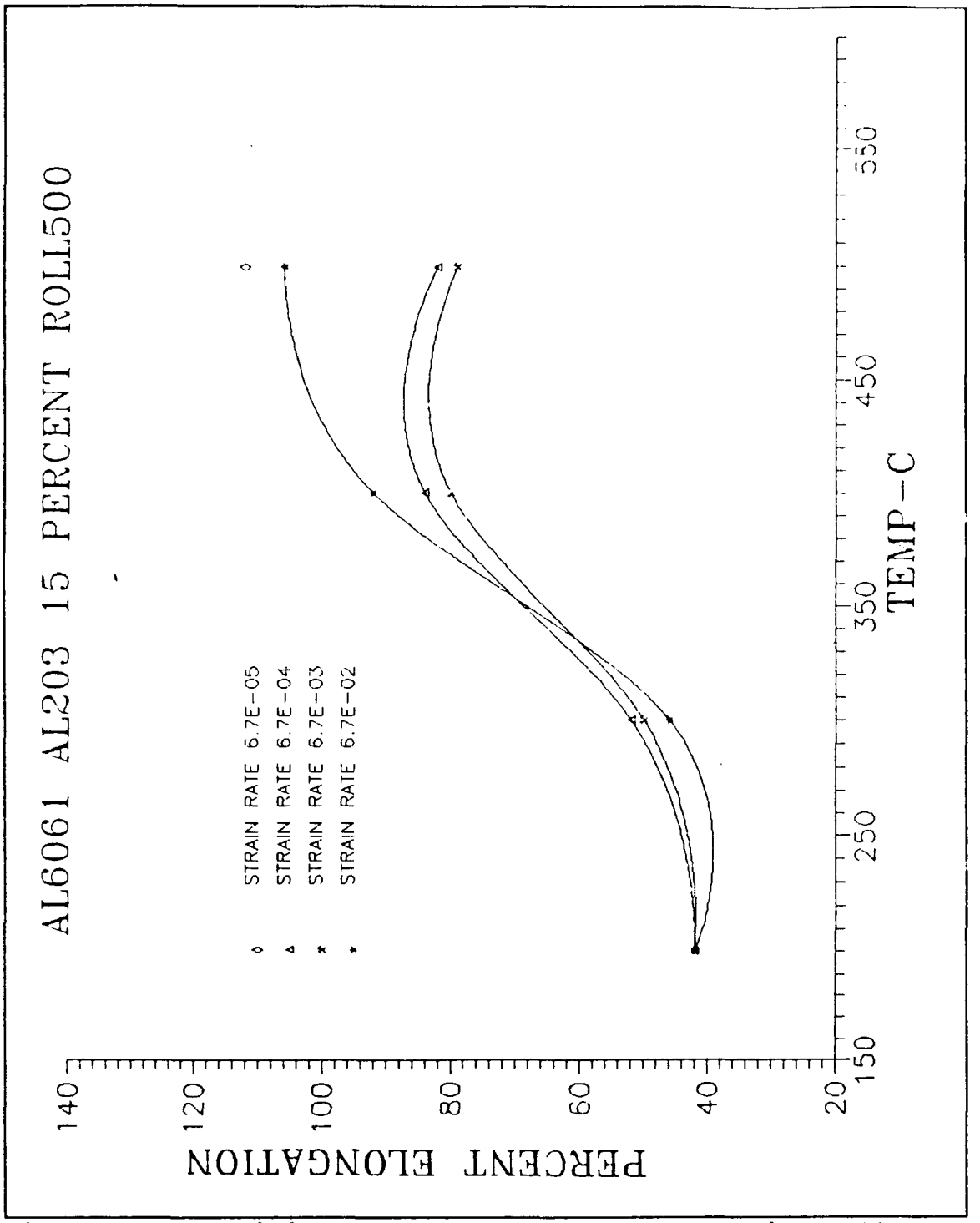

Figure 12 Ductility data for 15 vol. pct. material rolled at 500°C

V. DISCUSSION

The strength of composites is a function of the volume percentage and distribution of the reinforcement, the characteristics of the matrix, and the reinforcement-matrix interface. It has been shown here that the distribution of the reinforcement differs for the two rolling temperatures used. This may indicate that a lower rolling temperature will be more effective than a higher rolling temperature in redistribution of the alumina particles during processing. The alumina particles are relatively large, hard and non-deformable. During rolling deformation of a microstructure with an initially inhomogeneous particle distribution, a high dislocation density will be generated in the vicinity of the particle clusters. Complex tangles will form and this results in local strain hardening of the matrix [Ref. 17]. At locations distant from the clusters, the dislocation density will be lower and there will be less interaction among the dislocations and thus the material is weaker. As deformation proceeds, these weaker areas will deform more readily in comparison to the stronger regions near particle clusters. This will result in redistribution of the clusters in the microstructure. Similar considerations may explain the eventual break-up of the clusters to achieve a uniform particle distribution.

The development of a microstructure during processing by rolling schedules such as used in this study involves a balance between dislocation generation and recovery. For the particle size

of the alumina addition examined in this work, it is likely that dislocation structures generated during rolling deformation correspond to rotated regions in a finely structured deformation zone in the vicinity of the particles [Ref. 18]. With sufficient annealing time during processing, such structures may evolve into fine structures with boundaries of moderate misorientation angle. Such structures may strengthen the material at low temperature. At elevated testing temperatures, weakening may be observed, which reflects accelerated recovery associated with the fine structure. Since the tensile ductility of these materials is enhanced by the processing, it is unlikely that the apparent weakening of the processed composites is due to fracturing of the particles or damage to particle-matrix interfaces.

Similar considerations of microstructural evolution during processing may help in understanding the influence of processing on the temperature dependence of the ductility of these materials. At lower rolling temperatures, higher dislocation densities are achieved and thus the misorientation of boundaries evolved during annealing are greater within the deformation zones around the particles. Thus, materials rolled at the lower (350°C) temperature exhibit increased ductility at moderate test temperatures in comparison to those rolled at the higher (500°C) temperature.

For the composites rolled at 350°C, the decrease in ductility with increasing of the test temperature above the rolling temperature may be the result of grain growth prior to and during straining. If the particle to matrix interface bond is strong, and

deformation occurs rather than void formation, there will be a smaller grain size in regions adjacent to the particles. Grain growth in this region would therefore be rapid upon heating above the rolling temperature. It has also been noted that the matrix allloy exhibits precipitation at 100°C which results in formation of needle-like particles of a second phase [Ref. 19]. The role of such reactions may be a subject of future study.

It has also been noted in work on cold-rolled Copper-SiO_2 that the number of both primary and secondary dislocation loops near particles drops significantly with increasing deformation temperature [Ref. 17: pp167]. This results in a decrease in the rate of work hardening, as Orowan loops shrink and disappear by climb around particles. Local misorientations between small subgrains were noted in the vicinity of the particles in a manner similar to the mechanism proposed here.

Finally, in work on deformation mechanisms in Al6061-SiC, recovery mechanisms have been noted at a high temperature and at lower strain rates. However, recrystallization was observed at high strain rates and the same temperature [Ref. 20].

VI. CONCLUSIONS

1. The deformation characteristics of these DMMC materials tested are strain-rate and temperature sensitive. Thermomechanical processing by rolling at 350°C and 500°C altered to a limited degree the strength vs. temperature relationship and to a significant degree the ductility vs. temperature dependence of the composite from that of the extruded material.

2. Strength:

a. The strength vs. temperature dependence of the materials was essentially the same for the two rolling temperatures.

b. The 15 vol. pct. alumina material displayed higher strength than the 10 vol. pct. material for test temperatures up to 350°C. Above that temperature, strengths were identical.

c. The flow stress values for both the 10 vol pct. and 15 vol. pct. materials decreased with increasing temperature and decreasing strain rate.

d. Strengths obtained for rolled material were consistently slightly below those reported for those same composite materials in as-extruded or the -T6 temper condition.

3. Ductility:

a. In all cases the ductilities obtained were temperature and strain-rate dependent. Each material appeared to have maximum ductilities at or slightly above the prior rolling temperature.

b. The ductilities for both the 10 vol. pct. and 15 vol. pct. materials were very similar. A slightly greater ductility was noted for the 10 vol. pct. material rolled 350°C, when compared to the 15 vol. pct. material rolled at 350°C.

c. The materials both displayed a decrease in ductility with increasing strain rate for the test temperatures in the interval of 200-350°C. However, above this temperature, there is an inversion behavior and for temperatures in the range 400-500°C there is an increase in ductility associated with increasing strain rate.

d. For the 10 vol. pct. and 15 vol. pct. materials rolled at 350°C, a maximum in ductility of about 150 pct. was observed when testing was conducted at 400°C with a strain rate of $6.7x10_{-6}5^{-1}$.

e. At a temperature of approximately 410°C, a maximum in ductility as a function of strain rate was observed. In both 10 and 15 vol. pct. materials, ductilities for a stain rate of $3.3x10^{-1}S^{-1}$ were below those values obtained for a strain rate of $6.7x10^{-2}S^{-1}$.

VII. RECOMMENDATIONS FOR FURTHER STUDY

1. Investigate elevated temperatures properties for material rolled at 250°C.

2. Extend testing downward in temperature.

3. Devise schemes to vary the rolling schedule (reductions per pass, reheating, internal and total strain) as well as temperature.

4. Conduct transmission electron microscopy on the material to compare cast, extruded, and rolled materials and to assess differences in grain and subgain structures arising for the various rolling processes.

5. Conduct X-ray texture studies on the rolled and extruded materials.

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